

Development and Study of Hard-Facing Materials on the Base of Heat-Resisting High-Hardness Steels for Plasma-Jet Hard-Facing in Shielding-Doping Nitrogen Atmosphere

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Abstract: The authors develop hard-facing materials on the base of heat-resisting high-hardness steels for plasma-jet hard-facing in nitrogen atmosphere for manufacturing parts of mining and metallurgic equipment which significantly simplify the production process and effect a saving when producing bimetallic parts and tools.

Introduction

To strengthen machine parts for mining and metals industry the following chromium-tungsten heat-resisting high-speed steels are applied as hard-facing materials: R18, R6M5, R9, R2M8, 3H2V8, 3H2V4F, H10V14, 25H5FMS and other. Together with high performance properties these steels are characterized by insufficient weldability. Normally, to prevent cold cracking the traditional hard-facing technology involves mandatory application of high-temperature preheating and additional heating with the heating temperature T_{heating} of 400-700 °C and retarded cooling of the item. The process involves formation of plastic products of austenite decomposition characterized by low hardness and wear resistance which in its turn requires complex heat treatment. In the process of heat treatment of a bimetallic product the properties of a heavily alloyed metal may not be taken the full advantage of to ensure its maximal hardness [1].

According to the developed methods of alloying with heat-resisting high-hardness steels the authors suggest application of kinetic plasticity effect ("superplasticity") and regulated low-temperature heating to prevent cold cracking. The hard-facing methods allow producing built-up metal which properties are close to those of the quenched metal and which does not have cracks. The suggested methods of hard-facing can be implemented with application of reversed polarity plasma-jet hard-facing in shielding-doping nitrogen atmosphere which requires special materials [2-4].

Objective of the work is development and studying of hard-facing materials on the base of heat-resisting high-hardness steels for reverse polarity plasma-jet hard-facing in the shielding-doping nitrogen atmosphere parts of mining and metallurgic equipment.

Metal resistance to cold cracking may be changed to a certain extent regulating the residual welding stress increasing in the metal being cooled. The easiest and the most reliable method of such



regulation is to choose the chemical composition of the hard-faced metal which determines the linear expansion coefficient, character and structure and bulk effect of structural changes. These factors have a considerable influence upon residual stress development in the process of hard-facing. When studying the kinetic plasticity effect of R18 steel it was noted that the doping level has a certain impact over temporary stresses relaxation. Assumptions were made on the possibility of regulation of residual stresses by shifting the curve of temporary stresses formation into the region of lower temperatures due to proper alloying of the hard-faced metal. Then, temporary stresses in the austenite area which have been accumulated during metal cooling are mainly relaxed in the interval of martensite transformation and do not develop under cooling to room temperature T_r [2].

Below we provide the results of the study of how hard-facing impacts temporary stresses formation in the hard-faced metal to find the ways to reduce residual stress and, thus, to reduce the probability of cold cracking. The experiments were carried out with samples produced from hard-faced metal of 3H2V8 type which was doped with carbon, chromium, tungsten, aluminium by changing the chemical composition of the flux-cored wire. The impact of hard-facing over temporary and residual stress formation during cooling the samples produced from heat-resisting steels was studied with thermal microscopy unit IMASH-5S-69. Dumb bell samples with the section of 9 mm^2 and body of 46 mm long were produced from heat-resisting high-hardness steels. Effect of kinematic plasticity was estimated by temporary stresses reduction during cooling of rigidly held samples within the interval of martensite transformation. The samples were held with electromechanic drive of the loading system of the unit. The samples were heated in vacuum $5 \times 10^{-5} \text{ mm hg}$ with the heat given out during electric current flow. The heating was completed up to 1200 C ensuring the high degree of carbide phase solution in austenite. After heating and holding the samples were rigidly fixed to keep their length unchanged during cooling. The heating and cooling conditions imitated the thermal cycle of plasma-jet hard-facing. Unchanged length determined the corresponding growth of elastic plastic deformation of the sample during the cooling process. Stresses in the samples were constantly measured with strain-gage sensors with accuracy of $\pm 1.5 \%$. The temperature was measured with platinum to platinum/rhodium thermocouple welded to the central section of the sample. Kinetics of temperature-induced temporary stresses formation in rigidly fixed samples was rerecorded with two-dimensional potentiometer of PDP-4 type.

In Figure 1 the authors present data on the changes of temporary thermal stresses in rigidly fixed samples produced from hard-faced metal of 3H2V8, 4H4V10Yu and R18Yu steels type. From the data in Figure 1 curve 1 we can see that when the samples produced from steel R18Yu are cooled starting from 1200°C kinetic plasticity effect is observed under phase transition revealing itself in reduction of temporary stresses within the range of martensite temperatures. Maximum stress on the curve of temporary stresses formation is $\sim 240^\circ\text{C}$, which is rather close to the temperature under which martensite transformation of R18 steel starts under its hardening from heating temperature $T_h = 1200^\circ\text{C}$, and minimum of 100°C corresponds to such temperatures lying over the temperature of the end of martensite transformation $M_{\text{conditional end of martensitic transformations}}$ ($M_{\text{conditional end of martensitic transformations}} > M_{\text{the end of the martensitic transformation}}$) when the transformation degree is large enough and the yield limit grows due to formation of strong "framing" from numerous martensite crystals and plasticity of the alloy begins to reduce. From the data in Figure 1 we can see that in the region of martensite transformation temporary stresses reduce by 1.5-2 times as compared to the stresses accumulated in austenite region. Under further cooling below $M_{\text{conditional end of martensitic transformations}}$ temperature the stresses increase up to 120-180 mPa due to some increase of martensite amount. The results of the study showed that in hardening tool steels increased plasticity effect is observed in the moment of martensite transformation. Residual stresses in the samples produced from metal of 3H2V8 type are 200 mPa as opposed to 100 mPa in the samples produced from metal of 4H4V10Yu type. It can be explained by shifting of the curve of temporary stresses formation in 4H4V10Yu alloy towards the region of lower temperatures. Such shifting results from additional doping of the hard-faced metal with carbon, chromium and aluminium. Further increase of carbon and tungsten concentration in the hard-faced metal of R18Yu type resulted in shifting $M_{\text{the beginning of the martensitic transformation}} - M_{\text{conditional end of martensitic}}$

transformations into the region of still lower temperatures. After relaxation in the region of martensite transformation development of temporary stresses was not significant and residual stress in hard-faced metal R18 appeared to be lower than that in the metal of 4H4V10Yu type.

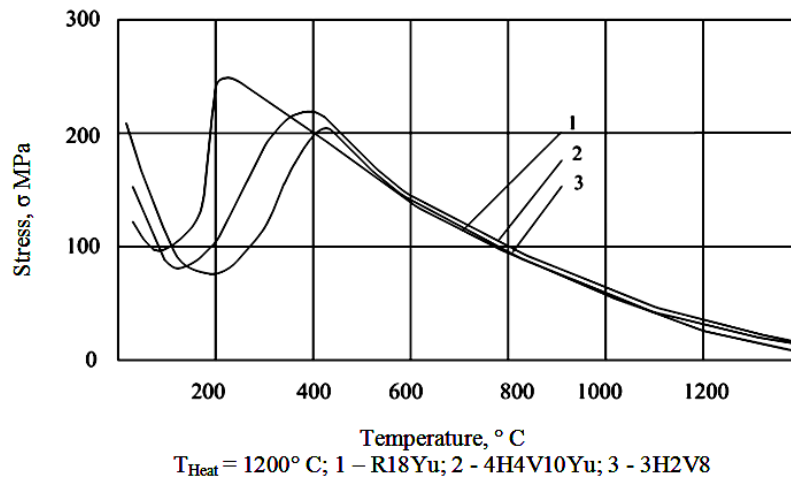


Figure 1. Effect of complex doping on temporary stress formation in the metal

Complex doping of hard-faced metal with carbon, chromium, tungsten and aluminium allows formation of finer austenite grains, so martensite in more doped hard-faced metal of R18Yu type is more disperse than martensite in 3H2V8 steel. The answer to the question of how completely we can use the martensite transformation as a mechanism of internal peak stresses relaxation depends upon the degree of martensite dispersion. The higher degree of relaxation of temporary stresses in steel R18 can be explained by the same phenomenon. It is worth noting that, due to the small value of temperature interval $M_{\text{conditional end of martensitic transformations}} - T_k$, temporary stresses cannot develop significantly and under the room temperature the lowest level of residual stress is noted in the samples produced from hard-faced metal R18Yu [2].

According to the completed study the hard-faced metal of R18Yu steel type is hard enough. So, after solving the problem of cold cracking prevention through additional doping R18Yu steel, this grade of steel may be recommended as a base for hard-facing materials for the given methods. To prove that proper doping of hard-faced metal can be used to regulate the residual stress level and, thus, the probability of cold cracks formation, hard-faced metal of R18Yu type was additionally doped with cobalt and nickel. In Figure 2 we present the data on how doping samples produced from R18Yu metal with cobalt effects shifting the martensite transformation interval and formation of temporary stresses. As we know doping with cobalt results in shifting martensite transformation into the region of higher temperatures. So if the hard-faced metal R18Yu does not contain cobalt, point M_n determined according to maximum on the curve (see Fig. 2) equals 180 °C; if metal is doped with 1% of cobalt the same point reaches 250°C and after doping the metal with 2% of cobalt M_n increases up to 340°C. We can also see that shifting $M_{\text{the beginning of the martensitic transformation}}$ and $M_{\text{conditional end of martensitic transformations}}$ points into the region of higher temperatures results in higher level of residual stress. Doping with non-carbide-forming nickel disimproves heat resistance to some extent due to induced lower temperature of $\gamma \rightarrow \alpha$ transformation. That is why nickel is introduced in small quantities (1-2%). As we can see from Figure 3 additional doping of R18Yu steel with nickel has led to shifting of temporary stresses formation curve into lower temperature region and significant reduction of residual stress and, thus, we can expect increased resistance of such metal to cold cracking [2-7].

The results of laboratory tests were checked by field tests of the studied grades of hard-faced metal for cold cracking resistance. Multiple-bead laser-jet hard-facing with low-temperature heating and forced cooling of 30HGSA steel workpieces was completed according to the developed method. The preheating temperature was chosen to be a bit higher than the temperature under which martensite

transformation of R18Yu steel starts ($T_{Heat} = 230 - 280^{\circ}C$) [2, 4]. When hard-facing with flux core wire doped with 2% of Co visual cracks were formed when facing the first turns of the first layer. Reducing cobalt to 1% allowed avoiding cracks only in the first layer. Three layers were faced with R18Yu wire without cracking. Doping with nickel allowed to avoid cracking under five-six-layer facing. At the same time doping with nickel significantly increases the amount and the stability of retained austenite, thus, complicating further tempering to some extent. That is why it is recommended to avoid introducing significant amounts of nickel when doping. One per cent of nickel is enough to prevent cracking of the faced metal.

Thus, it was shown that proper doping of the hard-faced metal may increase its resistance to cold cracking applying low-temperature preheating.

To complete plasma-jet hard-facing in the shielding-doping nitrogen atmosphere the scientists of Siberian State Industrial University developed flux core wires ensuring a wide range of chemical compositions and performance characteristics of the hard-faced metal.

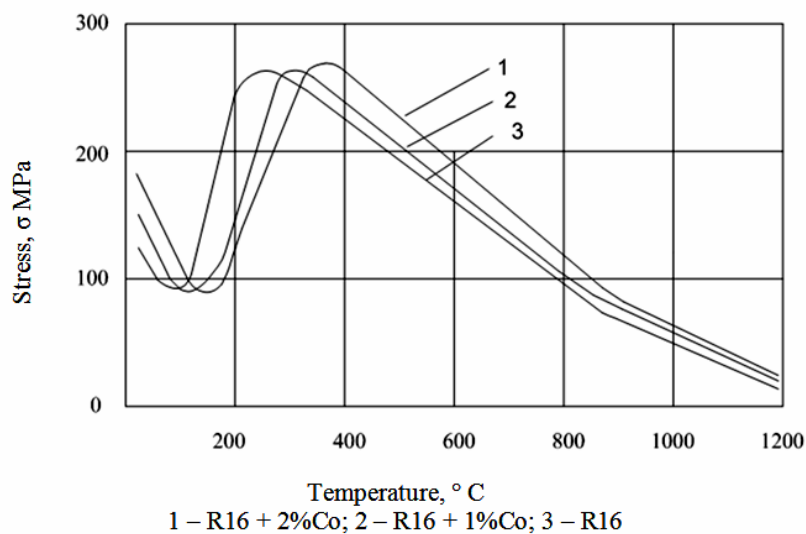


Figure 2. Effect of hard-faced metal of R18 type doping with cobalt upon the temporary stresses relaxation

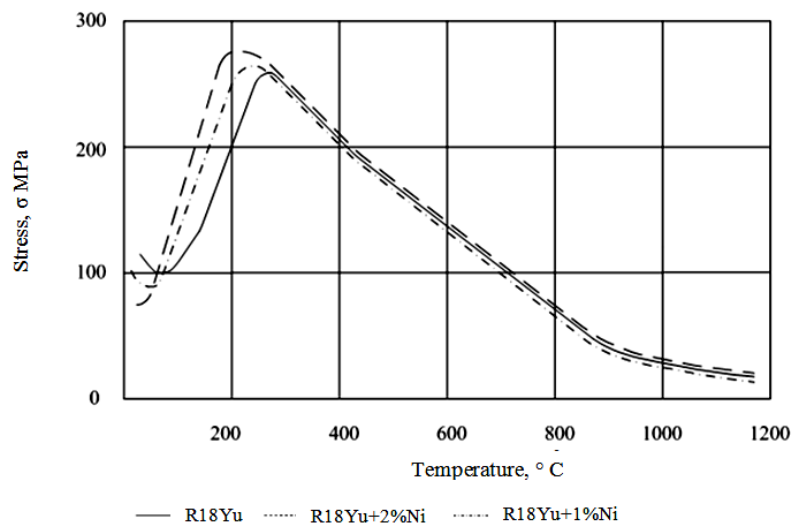


Figure 3. Effect of R18 steel doping with nickel upon the temporary stresses relaxation

Under hard-facing with constricted arc in nitrogen-containing gas atmospheres we observe intense boiling of the weld pool, as interaction between the molten metal of the weld pool and the gas phase in the process of hard-facing saturates the metal with nitrogen. When metal solidifies its ability to keep the dissolved nitrogen decreases abruptly and the excessive nitrogen escapes and, due to short duration of the hard-facing process, contributes to porosity formation. To avoid porosity aluminium was introduced into the core of the welding wire to fix the excessive nitrogen within compounds insoluble in the molten metal. Part of these compounds come to the surface of the weld pool and some stay in the hard-faced metal and can serve as the strengthening phase. On the other hand, aluminium transiting to the solid solution has a significant impact upon the process of structure formation as it increases the amount of soft structural components and, thus, contributes to destruction of the hard-faced metal.

After studying the process of structure formation of heat-resisting high-hardness tool steels hard-faced in nitrogen we established that aluminium affects metal hardness after hard-facing and tempering for secondary hardness. Aluminium participating in disperse particles formation, admittedly nitrides or compound oxides with the size of 6-12 nanometer contributes to metal strengthening but, if its amount is over 1%, it results in increased amount of δ -ferrite and softens the metal. For the purpose of obtaining maximal secondary hardness and porosity prevention optimal aluminium concentration in the hard-faced metal is between 0.8-1.0 % [2,7]. To reduce the ability for porosity formation caused by hydrogen, sodium fluorosilicate or fluoaluminic (Na_2SiF_6 , Na_3AlF_6) were additionally introduced into the core of flux cored wire electrode to fix hydrogen into compound [HF]. The possibility of applying nitrogen as shielding gas instead of argon allows reducing material costs, increase hard-facing productivity, additionally strengthen the hard-faced metal by doping it with nitrogen from the gas phase in the process of hard-facing. The hard-faced metal is not liable to cold cracking and its state is close to that of tempered metal allowing application of only high-temperature tempering to increase hardness and wear resistance.

The developed flux cored wire electrodes ensure producing hard-faced metal of high-hardness steels type such as R18, R6M5, R2M9, etc. Flux core wires were designed for plasma-jet hard-facing in the shielding-doping atmosphere of nitrogen for restoration and strengthening of wearing parts of equipment and metal-working tools when especially high hardness and wear resistance of effective areas is required. Diameter of the produced flux core wire is $2.7 + 3.7$ mm; metal deposit factor – 18–20 g/A·h; scattering coefficient does not exceed 3.5%. Hardness of the hard-faced metal HRC 52-60, hardness of the hard-faced metal after thermal treatment (high-temperature tempering) – HRC 62-66.

The state of the metal after hard-facing as well as after hard-facing and thermal treatment was estimated by the results of X-ray diffraction study and physicochemical phase analysis. Physicochemical analysis is completed to study the distribution of carbide-forming elements between the solid solution and the carbonitride phase. The samples were subjected to preliminary anodic dissolution on the surface to remove the layer changed after spark cutting and polishing. The physicochemical isolating of the carbonitride phase was completed in 15% water solution of sodium chlorate and 2.5% solution of tartaric acid. Current density was 0.03-0.06 A/cm², electrolyze time was 30-40 min. The amount of carbide-forming elements in the carbonitride phase was determined in relation to the weight of the dissolved metal. The phase composition of hard-faced high-speed steels of R18 (a, b) and R6M5 (c, d) types is shown in Fig. 4.

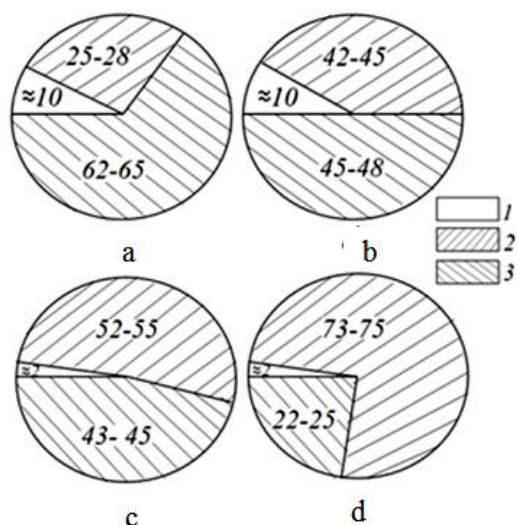


Figure 3. Phase composition of the hard-faced high-speed steels of R18 type(a, b) and R6M5 type (c, d): a, c – thermally untreated; b, d – after hard-facing and tempering; 1 – carbonitride $M_6(CN)$; 2 – martensite; 3 – retained austenite

In temperature-resistant steels carbides of M_6C , $M_{23}C_6$, M_2C , MC type can be formed, in steels containing nitrogen nitrides and carbonitrides of $M_6(CN)$ type can be also formed. X-ray diffraction study showed that the main constituent of anode deposit is the phase isomorphous to M_6C carbide. Anode deposit also contains small amount of $M_{23}C_6$ and traces of M_2C . The fact that hard-faced steels which have not been thermally treated have the form of solid solution doped with tungsten to a greater extent can be also explained by heating the doping metal in the arc as hard-facing to the temperatures significantly exceeding those of high-speed steel tempering. In this case heat resistance and secondary hardness of hard-faced high-speed steels exceed heat resistance and secondary hardness of tempered and thermally treated metal. Hard-faced high-speed steels of R18 and R6M5 types which have not been thermally treated contain the increased amount of retained austenite, their solid solution is more doped ($\alpha+\gamma$) and their hardness is lower in comparison to that of hard-faced thermally treated steels. The increased amount of retained austenite and higher doping of the solid solution of hard-faced steels can be explained by rapid cooling of the metal from the temperatures exceeding optimal temperatures of high-speed steels tempering. Hardness after high-temperature tempering of hard-faced high-speed steels equals or exceeds hardness of thermally treated tempered steels.

Industrial tests under real production conditions proved the results of full-scale and indirect study of the properties of hard-faced heat-resistant metal. It was established that the working rollers of a cold rolling mill produced with application of plasma-jet hard-facing with the developed alloying materials on the base of heat-resistant high-hardness steels with regulated thermal cycle and low-temperature heating meet the requirements under improving their strength by 1.5...2.0 times in comparison to series rolls. Improvement of wear resistance of hard-faced rolls can be explained by fine-disperse carbides M_6C и MC and nitrides and carbonitrides $M_6(CN)$ in their structure.

Conclusion

1. The authors develop flux cored wires for plasma-jet hard-facing parts of mining-metallurgic equipment in nitrogen atmosphere under reverse polarity. The charge material of the wires contains carbon, chromium, tungsten, molybdenum, aluminium, nickel and sodium fluorosilicate or fluoaluminic.

2. Complex doping of the hard-faced metal with carbon, chromium, tungsten, molybdenum, aluminium, nickel, sodium fluorosilicate or fluoaluminic allows producing hard-faced metal which is not liable to crack formation and which state is close to that of the tempered metal.

3. Hardness of the metal hard-faced with the developed materials is HRC 52–60, hardness of the hard-faced metal after the thermal treatment (tempering) is HRC 62 – 66. Thermal treatment of the hard-faced parts can be limited to tempering for secondary hardness without preliminary annealing and hardening. This significantly the technological process and effects a saving when producing bimetallic parts and tools.

References

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